

DESCRIPTION

HOT ROLLED STEEL SHEET FOR PROCESSING AND
METHOD FOR MANUFACTURING THE SAME

5

TECHNICAL FIELD

The present invention relates to a hot rolled steel sheet for processing having superior bake hardenability after aging, and a method for manufacturing the same.

The present application claims priority on Japanese Patent Application No.

10 2003-332013, filed on September 24, 2003, the content of which is incorporated herein by reference.

BACKGROUND ART

The use of light metals such as aluminum (Al) alloy and high-strength steel sheets
15 for automobile members has recently been promoted for the purpose of reducing weight in order to improve automobile fuel consumption. The light metals such as Al alloy offer the advantage of high specific strength; however, since they are much more expensive than steel, their applications are limited to special applications. Thus, there is a need to increase the strength of steel sheet to promote cost decreases and automobile weight
20 reductions over a wider range.

Since increasing the strength of a material typically causes deterioration of moldability (processability) and other material characteristics, the key to developing high-strength steel sheet is the extent to which strength can be increased without deteriorating material characteristics. Since characteristics such as burring formability,
25 ductility, fatigue durability and corrosion resistance are important characteristics that are

required of steel sheet used for inner plate members, structural members and underbody members, and how effectively these characteristics can be balanced with high strength on a high order is important.

For example, Japanese Unexamined Patent Application, First Publication Nos. 2000-169935 and 2000-169936 disclose transformation induced plasticity (TRIP) steel in which moldability (ductility and deep drawability) are dramatically improved as a result causing the occurrence of TRIP phenomenon during molding by containing residual austenite in the microstructure of the steel in order to achieve both high strength and various advantageous characteristics, especially moldability as previously described.

Steel sheet obtained in this art demonstrates breaking elongation in excess of 35% and superior deep drawability (limiting drawing ratio (LDR)) due to the occurrence of TRIP phenomenon by the residual austenite at a strength level of about 590 MPa.

However, amounts of elements such as C, Si and Mn must inevitably be reduced in order to obtain steel sheet having strength within the range of 370 to 540 MPa, and when the amounts of elements such as C, Si and Mn are reduced to realize the strength within the range of 370 to 540 MPa, there is the problem of being unable to maintain amount of residual austenite required for obtaining TRIP phenomenon in the microstructure at room temperature. Thus, it is difficult to apply high-strength steel sheet having strength of 540 MPa or higher to a member in which steel sheet having strength on the order of 270 to 340 MPa is currently used, without first improving operations and equipment used during pressing. The only realistic solution for the time being is to use steel sheet having strength of about 370 to 490 MPa. On the other hand, requirement for reduction of gauges is increasing year by year in order to achieve reduction in weight for automobile body, and it is therefore important for reduction in weight for automobile body to maintain pressed product strength as much as possible, based on the premise of reducing gauges.

Bake-hardening (BH) steel sheet has been proposed as a way of solving these problems because it has low strength during press molding and improves the strength of pressed products as a result of introducing stress due to pressing and subsequent baking finish treatment.

5 It is effective to increase solute C and solute N so as to improve bake hardenability; however, increases in these solute elements present in the solid solution worsen aging deterioration at normal temperatures. Consequently, it is important to develop a technology that can allow both bake hardenability and resistance to aging deterioration at normal temperatures.

10 On the basis of the requirements described above, Japanese Patent Application, Nos. H09-278697 and 2000-028141 disclose technologies for realizing both bake hardenability and resistance to aging deterioration at normal temperatures, in which bake hardenability is improved by increasing the amount of solute N, and the diffusion of solute C and solute N at normal temperatures is inhibited by an effect of increasing grain
15 boundary surface area caused by grain refining of crystal grains.

 However, the use of finer crystal grains has the risk of leading to increases in the yield point and causing deterioration of press moldability. In addition, increasing the amount of solute N offers the advantage of increasing the BH amount; however, there is concern over considerable decreases in the BH amount after aging due to the appearance
20 of yield point elongation caused by aging.

DISCLOSURE OF THE INVENTION

 The present invention relates to a hot rolled steel sheet for processing and a method for manufacturing the same, which has superior bake hardenability after aging
25 within a strength range of 370 to 490 MPa that allows to obtain a stable BH amount of 60

MPa or more since the hot rolled steel sheet has superior press moldability due to having a low yield ratio and there is little decrease in the BH amount due to aging. Namely, the present invention aims to provide a hot rolled steel sheet for processing having superior bake hardenability after aging that allows to stably manufacture pressed product having strength equivalent to that of pressed product manufactured by applying a 540 to 640 MPa-class steel sheet as a result of the introduction of pressing stress and baking finish treatment, even when the tensile strength of the hot rolled steel sheet is 370 to 490 MPa, and a method for manufacturing that steel sheet inexpensively and stably.

The inventors of the present invention conducted extensive research so as to obtain a steel sheet having superior bake hardenability after aging (little decrease in the BH amount caused by aging) as well as superior press moldability, with the emphasis on a production process for 370 to 490 MPa-class steel sheet produced on an industrial scale using ordinary production equipment currently in use.

As a result, the inventors of the present invention newly found that, a steel sheet in which C = 0.01 to 0.2%, Si = 0.01 to 0.3%, Mn = 0.1 to 1.5%, P ≤ 0.1%, S ≤ 0.03%, Al = 0.001 to 0.1% and N ≤ 0.006%, and as a remainder, Fe and unavoidable impurities is included, wherein the microstructure includes a main phase in the form of polygonal ferrite polygonal ferrite and a hard second phase, a volume fraction of the hard second phase is 3 to 20%, a hardness ratio (hardness of the hard second phase / hardness of the polygonal ferrite) is 1.5 to 6, and a grain size ratio (grain size of the polygonal ferrite / grain size of the hard second phase) is 1.5 or more, is extremely effective, thereby leading to completion of the present invention.

Namely, the gist of the present invention is as described below.

A hot rolled steel sheet of the present invention includes: in terms of percent by mass, C of 0.01 to 0.2%; Si of 0.01 to 0.3%; Mn of 0.1 to 1.5%; P of ≤0.1%; S of ≤0.03%;

Al of 0.001 to 0.1%; N of $\leq 0.006\%$; and as a remainder, Fe and unavoidable impurities, wherein the microstructure includes a main phase in the form of polygonal ferrite and a hard second phase, a volume fraction of the hard second phase is 3 to 20%, a hardness ratio (hardness of the hard second phase / hardness of the polygonal ferrite) is 1.5 to 6, and
5 a grain size ratio (grain size of the polygonal ferrite / grain size of the hard second phase) is 1.5 or more.

In accordance with the aforementioned aspect of the present invention, a hot rolled steel sheet for processing can be realized that has superior bake hardenability after aging. This hot rolled steel sheet has superior press moldability due to having a low
10 yield ratio, and also allows to obtain a stable BH amount of 60 MPa or more even in the case of having been exposed to an environment such that aging proceeds spontaneously after the steel sheet manufactured. Consequently, pressed product strength can be realized which is equivalent to that of pressed product manufactured by applying 540 to 640 MPa-class steel sheet, by introduction of pressing stress and baking finish treatment,
15 even when the steel sheet has tensile strength of 370 to 490 MPa. Therefore, the present invention can be said to have a high degree of industrial value.

In the aforementioned aspect, one or more selected from B of 0.0002 to 0.002%, Cu of 0.2 to 1.2%, Ni of 0.1 to 0.6%, Mo of 0.05 to 1%, V of 0.02 to 0.2%, and Cr of 0.01 to 1%, in terms of percent by mass, may be further included.

20 In the aforementioned aspect, one or two of Ca of 0.0005 to 0.005% and REM of 0.0005 to 0.02%, in terms of percent by mass, may be further included.

In the aforementioned aspect, the hot rolled steel sheet may be treated with zinc plating.

A method for manufacturing a hot rolled steel sheet for processing of the present
25 invention includes: a step of subjecting a slab having: in terms of percent by mass, C of

0.01 to 0.2%; Si of 0.01 to 0.3%; Mn of 0.1 to 1.5%; P of $\leq 0.1\%$; S of $\leq 0.03\%$; Al of 0.001 to 0.1%; N of $\leq 0.01\%$; and as a remainder, Fe and unavoidable impurities to a rough rolling so as to obtain a rough rolled bar; a step of subjecting the rough rolled bar to a finish rolling so as to obtain a rolled steel under conditions in which a sum of reduction rates of the final stage and the stage prior thereto is 25% or more, the reduction rate of the final stage is 1 to 15%, and a finishing temperature is in a temperature range from A_{r3} transformation point temperature to (A_{r3} transformation point temperature + 100°C); and a step of holding the rolled steel in a temperature range from below the A_{r3} transformation point temperature to the A_{r1} transformation temperature or higher for 1 to 15 seconds and then cooling to 350°C at a cooling rate of $100^{\circ}\text{C}/\text{sec}$ or more so as to obtain a hot rolled steel sheet, and coiling the hot rolled steel sheet at a temperature of below 350°C .

In the aforementioned aspect, a starting temperature of the finish rolling may be set to (A_{r3} transformation point temperature + 250°C) or higher.

In the aforementioned aspect, the rough rolled bar or the rolled steel may be heated during the time until the start of the step of subjecting the rough rolled bar to the finish rolling and/or during the step of subjecting the rough rolled bar to the finish rolling.

In the aforementioned aspect, descaling may be carried out during the time from the end of the step of subjecting the slab to the rough rolling to the start of the step of subjecting the rough rolled bar to the finish rolling.

In the aforementioned aspect, the resulting hot rolled steel sheet may be immersed in a zinc plating bath so as to galvanize the surface of the hot rolled steel sheet.

In the aforementioned aspect, an alloying treatment may be carried out after galvanizing.

FIG 1 is a graph in which the hardness ratio of steel sheet samples is plot against the volume fraction of the hard second phase.

BEST MODE FOR CARRYING OUT THE INVENTION

5 The following provides an explanation of the results of basic research leading to the present invention.

The following experiment was conducted to investigate the relationship between bake hardenability after aging and steel sheet microstructure. Slabs having the steel components shown in Table 1 were melted to prepare steel sheets having a thickness of 2 mm produced in various production processes, and then their bake hardenability after aging and their microstructure were examined.

Table 1

(% by mass)

C	Si	Mn	P	S	Al	N
0.068	0.061	1.22	0.009	0.001	0.034	0.0029

15

Bake hardenability after aging was evaluated in accordance with the following procedure. No. 5 test pieces as described in JIS Z 2201 were cut out of each steel sheet, and the test pieces were subjected to artificial aging treatment for 60 minutes at 100°C. Furthermore, preliminary tensile strain of 2% was applied to the test pieces, and then the test pieces were subjected to heat treatment equivalent to a baking finish treatment at 170°C for 20 minutes, after which the tensile test was carried out again. The tensile test was carried out in accordance with the method of JIS Z 2241.

Here, superior bake hardenability after aging indicates a large BH amount after

artificial aging treatment. In addition, the BH amount is defined as the value obtained by subtracting a flow stress of the preliminary tensile strain of 2% from the upper yield point obtained in the repeated tensile test.

On the other hand, microstructure was investigated in accordance with the following method. Samples cut out from a location of $1/4W$ or $3/4W$ of the width (W) of the steel sheets were ground along the cross-section in the direction of rolling, and then were etched using a nital reagent. Photographs were taken of the fields at $1/4t$ and $1/2t$ of the sheet thickness (t) and at a depth of 0.2 mm below a surface layer at 200-fold to 500-fold magnification using a light microscope.

Volume fraction of the microstructure is defined as the surface fraction in the aforementioned photographs of the metal structure. Next, measurement of average crystal grain sizes of the polygonal ferrite and the second phase was carried out using the comparison method described in JIS G 0552. Value, m of the crystal grains per 1 mm^2 of the cross-sectional area was calculated from the grain size number G determined from the measured values obtained by that comparison method using the equation of $m = 8 \times 2^G$. And then, the average crystal grain size d_m obtained from this value of m using the equation of $d_m = 1/\sqrt{m}$ is defined as the average crystal grain size of the polygonal ferrite and the second phase.

Here, for measuring the average crystal grain size, a method in which the aforementioned images observed using a light microscope are scanned to an image processing apparatus and so forth and equivalent circular diameter is calculated to use as the average crystal grain size, may also be used. The grain size ratio of the main phase in the form of polygonal ferrite to the second phase is defined as the average crystal grain size of the polygonal ferrite (d_m)/ average crystal grain size of the second phase (d_s).

Moreover, the hardness ratio of the hard second phase to the main phase in the

form of polygonal ferrite is defined as the Vickers hardness of the hard second phase (HV(s))/Vickers hardness of the main phase (Hv(m)). The Vickers hardness values of the hard second phase and the main phase are the average values obtained by measuring at least 10 points each in accordance with the method described in JIS Z 2244 and taking the average of values in which their respective maximum and minimum values are excluded.

The BH amount after aging, volume fraction of the second phase, and hardness ratio were measured in accordance with the methods described above, and the results are shown in FIG. 1. In the graph, steel sheets in which the volume fraction of the hard second phase is 3 to 20% and the hardness ratio is 1.5 to 6 are plotted with circles, while other steel sheets are plotted with squares. In addition, the BH amounts after aging of the steel sheets are indicated as numerical values inside the plotted points of those steel sheets.

The microstructures of the steel sheets are described near the plotted points. In FIG. 1, PF indicates polygonal ferrite, BF indicates bentonitic ferrite, M indicates martensite, B indicates bainite and P indicates pearlite.

As shown in FIG. 1, BH amount after aging, the volume fraction of the second phase, and the hardness ratio demonstrate an extremely strong correlation, and it was newly found that the BH amount after aging is 60 MPa or more in the case in which the volume fraction of the second phase is 3 to 20% and the hardness ratio is 1.5 to 6.

This mechanism is not completely understood; however, in the case in which a hard second phase is included in the microstructure in the optimum state (as for such as volume fraction and hardness ratio), numerous mobile dislocations are introduced as a result of the second hard phase undergoing a transformation at low temperatures at the time of production. If these mobile dislocations are introduced to a certain degree, it is presumed that the occurrence of yield point elongation and increases in the yield point are inhibited even after aging, and strain caused by processing is effectively reflected in the

BH amount.

The following provides a more detailed explanation of the microstructure of a steel sheet in the present invention.

In the present invention, it is necessary that the microstructure includes polygonal
5 ferrite and a hard second phase, and the hard second phase is either martensite or bainite.

In the case in which the hard second phase is martensite, since martensite has greater volumetric expansion and allows the introduction of a larger number of mobile dislocations than bainite, the yield point can be further lowered and the BH amount can be increased. Therefore, the hard second phase is preferably martensite. However,

10 residual austenite is allowed up to about 3%, which is the level at which it is unavoidably contained.

As previously described, it is required that the volume fraction of the second phase is 3 to 20% and the hardness ratio is 1.5 to 6 in order to realize both processability and superior bake hardenability after aging.

15 In the case in which the hard second phase is less than 3%, sufficient amount of mobile dislocations for inhibiting occurrence of yield point elongation even after aging and preventing lowering of the BH amount, cannot be obtained, while in the case in which the hard second phase exceeds 20%, the volume fraction of the main phase in the form of polygonal ferrite decreases, resulting in deterioration of processability. Thus, in order to
20 obtain a high BH amount even after aging, the volume fraction of the second phase should be 3 to 20%.

In the case in which the hardness ratio of the hard second phase to the main phase in the form of polygonal ferrite is less than 1.5, sufficient amount of mobile dislocations cannot be obtained for inhibiting occurrence of yield point elongation even after aging and
25 preventing lowering of the BH amount, while in the case in which the hardness ratio

exceeds 6, the effects are saturated. Thus, the hardness ratio should be from 1.5 to 6.

On the other hand, the main phase is made to be polygonal ferrite in order to obtain superior processability, and in addition, in order to obtain this effect, it is necessary that the grain size ratio of the polygonal ferrite to the second phase is 1.5 or more. In the case in which the grain size ratio of the polygonal ferrite to the second phase is less than 1.5, ductility decreases due to the influence of the hard second phase. Furthermore, if the hard second phase is a phase in which dissolved elements are concentrated and hardness has increased in the manner of martensite, the grain size of the second phase inevitably tends to become smaller. Since this results in greater resistance to the effects of the hard second phase; thereby, ductility is improved, the crystal grain size is preferably 2.5 or more.

In addition, in the case in which the average grain size of the polygonal ferrite is greater than 8 μm , yield stress decreases; thereby, moldability is improved. Therefore, the average grain size is preferably greater than 8 μm . There is no particular mention of the upper limit of the average grain size of the polygonal ferrite; however, it is preferably 25 μm or less from the viewpoint of surface roughness and so forth.

Moreover, the maximum height R_y of the steel sheet surface is preferably 15 μm (15 μm R_y , l (standard length: sampling length) 2.5 mm, l_n (evaluation length: travelling length) 12.5 mm) or less. This is because, as is described, for example, on page 84 of the Metal Material Fatigue Design Handbook, Society of Materials Science, Japan, the fatigue strength of hot rolled or acid washed steel sheet is clearly correlated with the maximum height R_y of the steel sheet surface.

In the present invention, it should be noted that not only is the BH amount at the preliminary strain of 2% superior evaluated as previously described, but also the BH amount at the preliminary strain of 10% is 40 MPa or more even in the case of $N \leq$

0.006%, and an amount of increase in tensile strength (ΔTS) at the preliminary strain of 10% is 40 MPa or more.

The following provides an explanation of the reason for limiting the chemical components of the present invention.

5 In the case in which the content of C is less than 0.01%, adequate hardness and volume fraction for the second phase cannot be obtained for inhibiting aging deterioration, and also the amount of C that is able to be present in a state of solid solution in the steel sheet decreases, which results in the risk of causing a decrease in the BH amount. Therefore, the content of C should be 0.01% or more. In addition, in the case in which
10 the content of C is more than 0.2%, the volume fraction of the second phase increases; thereby, strength is increased, which in turn results in deterioration of processability. Therefore, the content of C should be 0.2% or less. Moreover, the content of C is preferably 0.1% or less in the case of requiring a certain degree of hole expandability.

 Si and Mn are important elements in the present invention. They are required to
15 be included at specific amounts in order to obtain the required compound structure which includes polygonal ferrite and the second phase of the present invention, despite having low strength of 490 MPa or less. Mn in particular has the effect of expanding the temperature range of the ferrite and austenite dual phase state during cooling after completion of rolling and facilitates the obtaining of the required compound structure
20 including polygonal ferrite and the second phase of the present invention. Therefore, Mn is included at a content of 0.1% or more. However, since the effect of Mn is saturated when included at a content of more than 1.5%, the upper limit is made to be 1.5%.

 On the other hand, since Si has the effect of inhibiting precipitation of iron carbides during cooling, Si is included at a content of 0.01% or more. However, if
25 included in excess of 0.3%, its effect acts excessively, which makes it difficult to obtain

the compound structure including polygonal ferrite and the second phase. Moreover, in the case in which the content of Si is more than 0.3%, there is the risk of causing deterioration of processability for phosphating. Therefore, the upper limit of the content of Si is preferably 0.3%. In addition, in the case in which elements other than Mn that inhibit occurrence of hot cracks due to S are not adequately included, Mn is preferably included so that the contents of Mn and S satisfy $Mn/S \geq 20$ in terms of percent by mass. Moreover, in the case in which Mn is included so that the contents of Si and Mn satisfy Si + Mn of more than 1.5%, strength becomes excessively high, and this causes deterioration of processability. Therefore, the upper limit of the content of Mn is preferably 1.5%.

P is an impurity and its content should be as low as possible. In the case in which the content of P is more than 0.1%, P causes negative effects on processability and weldability. Therefore, the content of P should be 0.1% or less. However, it is preferably 0.02% or less in consideration of hole expanding and weldability.

Since S not only causes cracking during hot rolling but also forms A type inclusions that cause deterioration of hole expanding if excessively large amount of S is present, the content of S should be made to be as low as possible. Allowable range for the content of S is 0.03% or less. However, in cases in which a certain degree of hole expanding is required, it is preferable that the content of S is 0.001% or less, and in cases in which a high degree of hole expanding is required, it is preferable that the content of S is 0.003% or less.

Al is required to be included at a content of 0.001% or more for the purpose of deoxidation of molten steel; however, its upper limit is made to be 0.1% since Al leads to increased costs. In addition, since Al causes increases in amount of non-metallic inclusions resulting in deterioration of elongation if excessively large amount of Al is included, it is preferable that the content of Al is 0.06% or less. Moreover, it is

preferable that the content of Al is 0.015% or less in order to increase the BH amount.

N is typically a preferable element for increasing the BH amount. However, since aging deterioration becomes considerable if N is included at a content of more than 0.006%, the content of N should be 0.006% or less. Moreover, in the case of being
5 premised on allowing to stand for two weeks or more at room temperature after production and then using for processing, the content of N is preferably added at 0.005% or less from the viewpoint of aging. In addition, the content of N is preferably less than 0.003% when considering allowing to stand at high temperatures during the summer or when exporting across the equator during transport by a marine vessel.

10 B improves quench hardenability, and is effective in facilitating the obtaining of the required compound structure including polygonal ferrite and the second phase of the present invention. Therefore, B is included if necessary. However, in the case in which the content of B is less than 0.0002%, the content is inadequate for obtaining that effect, while in the case in which the content of B is more than 0.002%, cracking of the slabs
15 occurs. Accordingly, the content of B is made to be from 0.0002% to 0.002%.

Moreover, for the purpose of imparting strength, any one or two or more of alloying elements for precipitation or alloying elements for solid solution may be included that are selected from Cu at a content of 0.2 to 1.2%, Ni at a content of 0.1 to 0.6%, Mo at a content of 0.05 to 1%, V at a content of 0.02 to 0.2% and Cr at a content of 0.01 to 1%.

20 In the case in which the contents of any of these elements are less than the aforementioned ranges, its effect is unable to be obtained. In the case in which their contents exceed the aforementioned ranges, the effect becomes saturated and there are no further increases in effects even if the contents are increased.

Ca and REM are elements which change forms of non-metallic inclusions acting
25 as origins of breakage and causing deterioration of processability, and then eliminate their

harmful effects. However, they are not effective if included at contents of less than 0.0005%, while their effects are saturated if Ca is included at a content of more than 0.005% or REM is included at a content of more than 0.02%. Consequently, Ca is preferably included at a content of 0.0005 to 0.005%, while REM is preferably included at a content of 0.0005 to 0.02%.

Here, steel having these for their main components may further include Ti, Nb, Zr, Sn, Co, Zn, W or Mg on condition that the total content of these elements is 1% or less. However, since there is the risk of Sn causing imperfections during hot rolling, the content of Sn is preferably 0.05% or less.

Next, the following provides a detailed description of the reason for limiting the method for manufacturing a hot rolled steel sheet of the present invention.

A hot rolled steel sheet of the present invention is manufactured by a method in which slabs are hot rolled after casting and then cooled, a method in which a rolled steel or hot rolled steel sheet after hot rolling is further subjected to heat treatment on a hot-dip coating line, or a method which further includes other surface treatment on these steel sheets.

The method for manufacturing a hot rolled steel sheet of the present invention is a method for subjecting a slab to a hot rolling so as to obtain a hot rolled steel sheet, and includes a rough rolling step of rolling the slab so as to obtain a rough rolled bar (also referred to as a sheet bar), a finish rolling step of rolling the rough rolled bar so as to obtain a rolled steel, and a cooling step of cooling the rolled steel so as to obtain the hot rolled steel sheet.

There are no particular limitations on the manufacturing method carried out prior to hot rolling, that is, a method for manufacturing a slab. For example, slabs may be manufactured by melting using a blast furnace, a converter or an electric arc furnace,

followed by conducting various types of secondary refining for adjusting the components so as to have the target component contents, and then casting using a method such as ordinary continuous casting, casting using the ingot method or thin slab casting. Scrap may be used for the raw material. In the case of using slabs obtained by the continuous
5 casting, hot cast slabs may be fed directly to a hot rolling machine, or the slabs may be hot rolled after cooling to room temperature and then reheating in a heating oven.

There are no particular limitations on the temperature for reheating the slabs; however, in the case in which the temperature is 1400°C or higher, the amount of scale removed becomes excessive, resulting in a decrease in yield. Therefore, the reheating
10 temperature is preferably lower than 1400°C. In addition, in the case of heating at a temperature of lower than 1000°C, operating efficiency is considerably impaired in terms of scheduling. Therefore, the reheating temperature for the slabs is preferably 1000°C or higher. Moreover, in the case of reheating at a temperature of lower than 1100°C, the amount of scale removed becomes small, thereby there is a possibility that inclusions in
15 the surface layer of the slab can not be removed together with the scales by subsequent descaling. Therefore, the reheating temperature for the slabs is preferably 1100°C or higher.

The hot rolling step includes a rough rolling step and a finish rolling step carried out after completion of that rough rolling, and a starting temperature of finish rolling is
20 preferably (A_{r3} transformation point temperature + 250°C) or higher, in order to inhibit material variations in the direction of sheet thickness. The upper limit of the starting temperature of finish rolling is not particularly specified; however, in the case in which the temperature exceeds 1250°C, there is the risk that the finishing temperature at completion of finish rolling exceeds (A_{r3} transformation point temperature + 250°C). Therefore, the
25 starting temperature of finish rolling is preferably 1250°C or lower. In order to make the

starting temperature of finish rolling equal to or higher than (Ar_3 transformation point temperature + 250°C), the rough rolled bar or the rolled steel is heated during the time from the end of the rough rolling to the start of the finish rolling and/or during the finish rolling, as necessary.

5 In order to obtain stable and superior breaking elongation in particular in the present invention, it is effective to inhibit the fine precipitation of MnS and so forth. Normally, precipitates such as MnS are redissolved in a solid solution during reheating of the slabs at about 1250°C, and finely precipitate during subsequent hot rolling. Thus, ductility can be improved by controlling the reheating temperature of the slabs to about
10 1150°C so as to prevent MnS from being redissolved in the solid solution. However, in order to make the finishing temperature at completion of rough rolling to be within the range of the present invention, it is an effective means to heat the rough rolled bar or the rolled steel during the time from the end of rough rolling to the start of finish rolling and/or during finish rolling. Any type of system may be used for the heating apparatus in
15 this case; however, a transverse system is preferable since it enables heating uniformly in the direction of sheet thickness.

In the case of carrying out descaling during the time from the end of the rough rolling to the start of the finish rolling, it is preferable that collision pressure P (MPa) and flow rate L (liters/cm²) of high-pressure water on the surface of the steel sheet satisfy the
20 conditional expression of $P \times L \geq 0.0025$.

The collision pressure P of the high-pressure water on the surface of the steel sheet is described in the following manner (see "Iron and Steel", 1991, Vol. 77, No. 9, p. 1450).

$$P \text{ (MPa)} = 5.64 \times P_0 \times V/H^2$$

25 where,

P_0 (MPa): Liquid pressure

V (liters/min): Flow rate of liquid from nozzle

H (cm): Distance between surface of steel sheet and nozzle

Flow rate L is described in the following manner.

5
$$L \text{ (liters/cm}^2\text{)} = V / (W \times v)$$

where,

V (liters/min): Flow rate of liquid from nozzle

W (cm): Width of spraying liquid that contacts the surface of the steel sheet per nozzle

10 v (cm/min): Sheet transport speed

It is not particularly necessary to specify the upper limit of value of collision pressure $P \times$ flow rate L in order to obtain the effects of the present invention; however, the upper limit of the value of collision pressure $P \times$ flow rate L is preferably 0.02 or less, since excessive nozzle wear and other problems occur when the nozzle liquid flow rate is
15 increased.

As a result of descaling, scale can be removed from the surface such that the maximum height R_y of the steel sheet surface is 15 μm (15 μm R_y , l (standard length: sampling length) 2.5 mm, l_n (evaluation length: traveling length) 12.5 mm) or less. In addition, the subsequent finish rolling is preferably carried out within 5 seconds after the
20 descaling so as to prevent reformation of scale.

In addition, sheet bars may be joined between the rough rolling and the finish rolling, and the finish rolling may be carried out continuously. At that time, the rough rolled bar may be temporarily coiling into the shape of a coil, put in a cover having a warming function if necessary, and then joined after uncoiling.

25 It is necessary to suitably promote ferrite transformation after completion of

rolling in order to obtain the desired fractions of the microstructure and hardness ratio between the main phase and the second phase in this component system. Therefore, it is necessary that the finish rolling be carried out under conditions in which a sum of reduction rates of the final stage and the stage prior thereto is 25% or more. In the case in which the reduction rate of the final stage is less than 1%, the flatness of the steel sheet deteriorates, while in the case in which it exceeds 15%, ferrite transformation proceed too much; thereby, the desired microstructure in which the grain size ratio of the polygonal ferrite to the second stage is 2.5 or more is not obtained. Therefore, the reduction rate of the final stage should be 1 to 15%. An upper limit is not particularly provided for the sum of reduction rates of the final stage and the stage prior thereto; however, it is preferably 50% or less in consideration of equipment restrictions due to rolling reaction force.

Moreover, finishing temperature (FT) at completion of the finish rolling should be in a temperature range from A_{r3} transformation point temperature to (A_{r3} transformation point temperature + 100°C). Here the A_{r3} transformation point temperature is simply indicated with, for example, the relationship with the steel components in accordance with the following calculation formula.

Namely, $A_{r3} = 910 - 310 \times \%C + 25 \times \%Si - 80 \times \%Mneq$, where $Mneq = \%Mn + \%Cr + \%Cu + \%Mo + \%Ni/2 + 10(\%Nb - 0.02)$.

Or, in the case of including B, $Mneq = \%Mn + \%Cr + \%Cu + \%Mo + \%Ni/2 + 10(\%Nb - 0.02) + 1$.

Here, the parameters of %C, %Si, %Mn, %Cr, %Cu, %Mo, %Ni, and %Nb in the formula indicate the respective contents (mass %) of elements C, Si, Mn, Cr, Cu, Mo, Ni and Nb in the slabs.

In the case in which the finishing temperature (FT) at completion of finish rolling

is lower than the Ar_3 transformation point temperature, there is the possibility of $\alpha + \gamma$ two-phase-rolling; thereby, processed structure remains in the ferrite grains after rolling, resulting in the risk of deterioration of ductility. Therefore, FT is made to be equal to or higher than the Ar_3 transformation point temperature. In addition, in the case in which the finishing temperature (FT) at completion of finish rolling exceeds (Ar_3 transformation point temperature + 100°C), the strain which is caused by rolling and is required for ferrite transformation after completion of rolling, is alleviated by recrystallization of austenite; thereby, the target microstructure is not obtained at the end. Therefore, the finishing temperature (FT) at completion of finish rolling is (Ar_3 transformation point temperature + 100°C) or lower.

After completion of the finish rolling, the temperature is held for 1 to 15 seconds within the temperature range of two-phase of $\alpha + \gamma$ that is below the Ar_3 transformation point temperature and equal to or higher than the Ar_1 transformation temperature. In the case in which the duration of this holding is less than 1 second, phase separation of ferrite phase and austenite phase does not proceed sufficiently; thereby, the target microstructure is not obtained at the end. Here, the Ar_1 transformation temperature is simply indicated by, for example, the relationship with the steel components in accordance with the following calculation formula:

$$Ar_1 = 830 - 270 \times \%C - 90 \times \%M_{neq}$$

On the other hand, in the case in which the duration of that holding exceeds 15 seconds, not only is there the risk of being unable to obtain the target microstructure due to the formation of pearlite, but also the sheet passage rate decreases which results in a considerable reduction in productivity. Therefore, the time during which the steel sheet is held in that temperature range is 1 to 15 seconds. Cooling until the temperature reaches that held temperature, is not particularly specified; however, the steel sheet is

preferably cooled to this temperature range at a cooling rate of 20°C/sec or more so as to promote separation of α and γ phases. Next, after completion of holding at the above temperature, the steel sheet is cooled to 350°C at a cooling rate of 100°C/sec or more and then coiled at a temperature below 350°C. In the case in which the steel sheet is cooled at a cooling rate of less than 100°C/sec, pearlite ends up forming which prevents the obtaining of a second phase of sufficiently hard; thereby, the target microstructure cannot be obtained. Therefore, adequate bake hardenability is unable to be secured. Thus, the cooling rate is made to be 100°C or more. The effects of the present invention can be obtained without particularly specifying the upper limit of the cooling rate; however, since there is concern over warping of the sheet caused by thermal strain, it is preferably 200°C/s or less.

In the case of a coiling temperature of 350°C or higher, a hardness ratio of 1.5 to 6 which is necessary to obtain sufficient mobile dislocations for lowering the BH amount without causing yield point elongation after aging, is not achieved. Therefore, the coiling temperature is limited to lower than 350°C. Moreover, the coiling temperature is preferably 150°C or less from the viewpoint of resistance to aging deterioration. In addition, it is not particularly necessary to limit the lower limit of the coiling temperature; however, since there is concern over a defective appearance caused by the presence of rust if the coil remains wet for a long period of time, it is preferably 50°C or higher.

After completion of the hot rolling step, acid washing may be carried out if necessary, and then skinpass at a reduction rate of 10% or less, or cold rolling at a reduction rate of up to about 40% may be carried out either offline or inline.

Furthermore, skinpass rolling is preferably carried out at 0.1% to 0.2% so as to correct the shape of the steel sheet and to improve ductility due to introduction of mobile dislocations.

In order to subject hot rolled steel sheet after acid washing to zinc plating, hot rolled steel sheet may be immersed in a zinc plating bath and if necessary, subjected to alloying treatment.

5

EXAMPLES

The following provides a more detailed explanation of the present invention through its examples.

After steels A to K having the chemical components shown in Table 2 were melted using a converter and were subjected to continuous casting, they were either sent
10 directly to rough rolling or reheated prior to rough rolling, and then were subjected to rough rolling and finish rolling so as to make sheet thickness 1.2 to 5.5 mm, and were coiled. The chemical compositions shown in the table are indicated in percent by mass (mass%).

Table 2

Slab No.	Chemical Composition (unit: mass%)								
	C	Si	Mn	P	S	Al	N	Si+Mn	Other
X1	0.071	0.06	1.21	0.011	0.001	0.031	0.0026	1.27	
X2	0.048	0.22	0.72	0.010	0.001	0.033	0.0038	0.94	Cu:0.29%, Ni:0.12%
X3	0.074	0.07	1.01	0.011	0.001	0.028	0.0027	1.08	B:0.004%, Cr:0.08%
X4	0.051	0.04	0.98	0.009	0.001	0.031	0.0029	1.02	Mo:0.11%
X5	0.072	0.05	1.08	0.009	0.001	0.016	0.0030	1.13	V:0.08%
X6	0.066	0.05	1.23	0.008	0.001	0.024	0.0028	1.28	REM:0.0009%
X7	0.063	0.04	1.31	0.010	0.001	0.026	0.0024	1.35	Ca:0.0014%
X8	0.052	0.03	1.02	0.010	0.001	0.034	0.0038	1.05	Cr:0.61%
Y1	0.070	<u>1.02</u>	0.36	0.008	0.001	0.035	0.0041	1.38	
Y2	0.070	0.03	1.26	0.012	0.001	0.015	<u>0.0084</u>	1.29	
Y3	<u>0.210</u>	<u>1.51</u>	1.49	0.010	0.001	0.033	0.0036	<u>3.00</u>	
Y4	0.064	<u>0.89</u>	1.26	0.010	0.001	0.034	0.0038	<u>2.15</u>	

The details of the production conditions are shown in Table 3. Here, "heating rough rolled bar" indicates heating of the rough rolled bar or the rolled steel during the time from the end of rough rolling to the start of finish rolling and/or during finish rolling, and indicates whether or not this heating has been carried out. "FT" indicates the finishing temperature at completion of finish rolling, "Holding time" indicates the air-cooling time in the temperature range from below the A_{r3} transformation point

temperature to equal to or higher than the Ar_1 transformation temperature, "Cooling rate from holding temperature range to 350°C" indicates the average cooling rate when the rolled steels were cooled in the temperature range from the holding temperature range to 350°C, and "CT" indicates the coiling temperature. Here, "MT" indicates the

- 5 temperature measured using a runout table intermediate thermometer, it is equivalent to the temperature at which cooling is started during "cooling from the holding temperature range to 350°C" in the examples.

- As shown in Table 3, descaling was carried out in Example 3 under conditions of a collision pressure of 2.7 MPa and flow rate of 0.001 liters/cm² after rough rolling. In
10 addition, zinc plating was carried out in Example 8.

Table 3-1

	Production Conditions				
	Slab No.	Heating rough rolled bar	Temperature at start of finish rolling (°C)	Ar ₃ +250 (°C)	(Reduction rate of final stage)/(Sum of reduction rates of final stage and stage prior thereto) (%/%)
Ex.1	X1	Yes	1100	1044	14/36
Ex.2	X1	No	1100	1044	14/36
Ex.3	X2	No	1100	1094	10/24
Ex.4	X3	No	1100	1059	10/24
Ex.5	X4	No	1100	1068	10/24
Ex.6	X5	Yes	1100	1053	10/24
Ex.7	X6	Yes	1100	1043	14/36
Ex.8	X7	Yes	1100	1038	14/36
Ex.9	X1	Yes	<u>980</u>	1044	14/36
Ex.10	X1	Yes	1000	1044	14/36
Ex.11	X7	Yes	1100	1038	14/36
Ex.12	X8	Yes	1100	1015	14/36
Comp.Ex.1	X1	Yes	1100	1044	<u>16/22</u>
Comp.Ex.2	X1	Yes	1100	1044	14/36
Comp.Ex.3	X1	Yes	1100	1044	14/36
Comp.Ex.4	X1	Yes	1100	1044	14/36
Comp.Ex.5	X1	Yes	1100	1044	<u>18/36</u>
Comp.Ex.6	Y1	No	1100	1136	14/36
Comp.Ex.7	Y2	No	1100	1038	14/36
Comp.Ex.8	Y3	Yes	1100	1014	10/26

Table 3-1 (Continued)

	Production Conditions								Comments
	FT (°C)	Ar ₃ (°C)	Ar ₃ +100 (°C)	Ar ₁ (°C)	MT# (°C)	Holding time (sec)	Cooling rate from holding temp. to 350°C (°C/sec)	CT (°C)	
Ex.1	850	794	894	712	720	4.0	120	<150	
Ex.2	850	794	894	712	720	4.0	120	<150	
Ex.3	870	844	944	722	740	5.0	110	200	*1
Ex.4	870	809	909	714	720	5.0	110	200	
Ex.5	870	818	918	714	730	5.0	110	200	
Ex.6	870	803	903	713	730	5.0	110	200	
Ex.7	850	793	893	711	720	5.0	110	200	
Ex.8	850	788	888	710	720	5.0	110	200	*2
Ex.9	850	794	894	702	710	4.0	120	<150	
Ex.10	850	794	894	702	710	4.0	120	<150	
Ex.11	850	788	888	710	740	1.5	100	250	
Ex.12	850	765	865	706	720	4.0	100	<150	
Comp.Ex.1	850	794	894	712	780	4.0	120	<150	
Comp.Ex.2	<u>780</u>	794	894	712	720	4.0	120	<150	
Comp.Ex.3	850	794	894	712	780	<u>0.5</u>	120	<150	
Comp.Ex.4	850	794	894	712	720	4.0	<u>10</u>	<u>500</u>	
Comp.Ex.5	850	794	894	702	710	4.0	120	<150	
Comp.Ex.6	890	886	986	749	750	4.0	120	<150	
Comp.Ex.7	850	788	888	710	720	4.0	120	<150	
Comp.Ex.8	875	764	864	751	760	5.0	110	<u>400</u>	

*1: Descaling was carried out after rough rolling under conditions of a collision pressure of 2.7 MPa and a flow rate of 0.001 liters/cm².

*2: The sheet was passed through a zinc plating step.

Table 3-2

	Microstructure				Mechanical Properties			Bake Hardenability	
	Microstructure	Volume fraction of Second phase (%)	Hardness ratio	Crystal grain size ratio	YP (MPa)	TS (MPa)	E1 (%)	2%BH (MPa)	2%BH after artificial aging (MPa)
Ex.1	PF+M	10	3.7	2.7	295	461	36	79	78
Ex.2	PF+M	8	3.9	2.8	289	456	35	81	81
Ex.3	PF+B	13	2.9	2.9	288	416	35	68	66
Ex.4	PF+M	8	3.8	2.8	312	488	32	91	88
Ex.5	PF+B	12	3.2	2.9	290	442	34	80	77
Ex.6	PF+B	14	2.7	2.7	320	491	32	77	70
Ex.7	PF+M	9	3.8	3.0	320	460	35	88	86
Ex.8	PF+M	10	3.6	2.9	324	471	34	80	80
Ex.9	PF+M	9	3.8	2.8	293	470	34	71	65
Ex.10	PF+M	6	4.1	2.9	297	460	33	74	63
Ex.11	PF+B	16	2.6	1.9	316	466	33	61	62
Ex.12	PF+B	18	1.8	1.8	344	481	31	64	61
Comp.Ex.1	<u>BF</u>	<u>100</u>	<u>1.0</u>	--	322	456	33	58	56
Comp.Ex.2	<u>Processed F+M</u>	5	1.2	2.6	389	470	28	61	58
Comp.Ex.3	<u>BF</u>	<u>100</u>	<u>1.0</u>	--	318	460	31	60	55
Comp.Ex.4	<u>PF+P</u>	12	<u>1.4</u>	2.7	311	439	32	21	8
Comp.Ex.5	PF+B	5	4.0	<u>1.4</u>	320	460	31	55	<u>45</u>
Comp.Ex.6	PF+M	<u>2</u>	2.7	2.2	410	570	24	12	10
Comp.Ex.7	PF+M	11	3.6	2.6	303	465	34	76	36
Comp.Ex.8	<u>PF+B</u> <u>+13%γR</u>	31	2.1	1.8	566	794	33	46	43

In the table, γ R indicates residual austenite.

Thin steel sheets obtained in this manner were evaluated by tensile tests and BH tests after artificial aging in the same manner as the evaluation methods described in the section on the best mode for carrying out the invention. Moreover, the microstructures of the steel sheets were similarly investigated, and the average grain sizes of the polygonal ferrite and the second phase, and the hardness ratio of the hard second phase to the main phase that is the polygonal ferrite, were measured. These results are shown in Table 3.

The hot rolled steel sheets of Examples 1 to 12 included the predetermined amounts of steel components, their microstructures includes a main phase in the form of polygonal ferrite and a hard second phase, the volume fractions of the second phases were 3 to 20%, the hardness ratios were 1.5 to 6, and the grain size ratios were 1.5 or more. In these Examples 1 to 12, the BH amount after artificial aging exceeded 60 MPa, and the hot rolled steel sheets for processing were obtained that have superior bake hardenability after aging.

Comparative Examples 1 to 8 other than those described above were outside the scope of the present invention for the reasons described below.

In Comparative Example 1, since the reduction rate of the final stage and the sum of reduction rates of the final stage and the stage prior thereto were outside the range of claim 5 of the present invention, the target microstructure described in claim 1 could not be obtained; thereby, adequate BH amount after artificial aging was not realized.

In Comparative Example 2, since the finishing temperature (FT) at completion of finish rolling was outside the range of claim 5, the target microstructure described in claim 1 could not be obtained; thereby, adequate BH amount after artificial aging was not realized.

In Comparative Example 3, since the holding time was outside the range of claim 5, the target microstructure described in claim 1 could not be obtained; thereby, adequate

BH amount after artificial aging was not realized.

In Comparative Example 4, the cooling rate in a temperature range from the holding temperature to 350°C was outside the range of claim 5. In particular, since the cooling rate in a temperature range from the holding temperature to 350°C was less than 100°C/sec., pearlite was formed. Thus, the target microstructure described in claim 1 could not be obtained; thereby, adequate BH amount after artificial aging was not realized.

In Comparative Example 5, since the reduction rate of the final stage was outside the range of claim 5, the target microstructure described in claim 1 could not be obtained; thereby, adequate BH amount after artificial aging was not realized.

In Comparative Example 6, since the content of Si in the slab Y1 used was outside the range of claim 1, the target microstructure described in claim 1 could not be obtained; thereby, adequate BH amount after artificial aging was not realized.

In Comparative Example 7, the target microstructure described in claim 1 was obtained; however, since the content of N in the slab Y2 used was outside the range of claim 1, aging deterioration was excessive; thereby, adequate BH amount after artificial aging was not realized.

In Comparative Example 8, the contents of C and Si in the slab Y3 used were outside the range of claim 1, and the coiling temperature was outside the range of claim 6. Therefore, the target microstructure described in claim 1 could not be obtained.

INDUSTRIAL APPLICABILITY

Since this hot rolled steel sheet for processing is capable of demonstrating a stable BH amount of 60 MPa or more due to the small amount of the decrease in the BH amount caused by aging, pressed product strength can be obtained which is equivalent to that of pressed product manufactured by applying steel sheets having tensile strength of

540 to 640 MPa, as a result of introduction of pressing stress and baking finish treatment, even when the tensile strength of the hot rolled steel sheet is 370 to 490 MPa.

Consequently, this hot rolled steel sheet for processing can be preferably used as steel sheet for industrial products to which reduction of gauges are strongly required for the purpose of achieving weight saving, as in the case of chassis parts and so forth of automobiles in particular.